# NATIONAL ADVISORY COMMITTEE FOR AERONAUTICS

REPORT No. 659

# EFFECT OF SERVICE STRESSES ON IMPACT RESISTANCE, X-RAY DIFFRACTION PATTERNS, AND MICROSTRUCTURE OF 25S ALUMINUM ALLOY

By J. A. KIES and G. W. QUICK



1939

#### **AERONAUTIC SYMBOLS**

#### 1. FUNDAMENTAL AND DERIVED UNITS

		Metric		English		
	Symbol	Unit	Abbrevia- tion	Unit	Abbrevia- tion	
Length Time Force	l t F	metersecondweight of 1 kilogram	m s kg	foot (or mile) second (or hour) weight of 1 pound	ft. (or mi.) sec. (or hr.) lb.	
PowerSpeed	P V	horsepower (metric) (kilometers per hour (meters per second	k.p.h. m.p.s.	horsepower miles per hour feet per second	hp. m.p.h. f.p.s.	

#### 2. GENERAL SYMBOLS

W, $g$ ,	Weight=mg Standard acceleration of gravity=9.80665 m/s <sup>2</sup> or 32.1740 ft./sec. <sup>2</sup>	ν, Kinematic viscosity ρ, Density (mass per unit volume) Standard density of dry air, 0.12497 kg-m <sup>-4</sup> -s <sup>2</sup>	at
m,	$\text{Mass} = \frac{W}{q}$	15° C. and 760 mm; or 0.002378 lbft. <sup>-4</sup> sec. <sup>2</sup> Specific weight of "standard" air, 1.2255 kg/m <sup>3</sup>	
I,	Moment of inertia= $mk^2$ . (Indicate axis of radius of gyration $k$ by proper subscript.)	0.07651 lb./eu. ft.	
μ,	Coefficient of viscosity		

#### 3. AERODYNAMIC SYMBOLS

R,

Resultant force

S,	Area	iw,	Angle of setting of wings (relative to thrust
$S_w$ ,	Area of wing	i,	line) Angle of stabilizer setting (relative to thrust
G, b,	Gap Span	011	line)
c,	Chord	Q,	Resultant moment
$\frac{b^2}{\overline{S}}$ ,		Ω,	Resultant angular velocity
	Aspect ratio	$\rho \frac{Vl}{}$	Reynolds Number, where $l$ is a linear dimension
V,	True air speed	$\rho \overline{\mu}$	(e.g., for a model airfoil 3 in. chord, 100
q,	Dynamic pressure $=\frac{1}{2}\rho V^2$		m.p.h. normal pressure at 15° C., the cor-
			responding number is 234,000; or for a model
L,	Lift, absolute coefficient $C_L = \frac{L}{gS}$		of 10 cm chord, 40 m.p.s., the corresponding
D	Drog absolute coefficient $C = D$	$C_p$ ,	number is 274,000) Center-of-pressure coefficient (ratio of distance
D,	Drag, absolute coefficient $C_D = \frac{D}{qS}$	Op,	of c.p. from leading edge to chord length)
$D_0$ ,	Profile drag, absolute coefficient $C_{D_0} = \frac{D_0}{qS}$	α,	Angle of attack
		€,	Angle of downwash
$D_i$ ,	Induced drag, absolute coefficient $C_{D_i} = \frac{D_i}{qS}$	$\alpha_0$ ,	Angle of attack, infinite aspect ratio
-		$\alpha_i$ ,	Angle of attack, induced
$D_{p}$	Parasite drag, absolute coefficient $C_{D_p} = \frac{D_p}{qS}$	$\alpha_a$ ,	Angle of attack, absolute (measured from zero-
C,	Cross-wind force, absolute coefficient $C_c = \frac{C}{qS}$	~	lift position) Flight-path angle
0,	Cross while force, absolute coefficient or qS	γ,	Fright-paul anglo

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By J. A. KIES and G. W. QUICK National Bureau of Standards

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#### SUMMARY

A great number of tests were made to determine the effect of service stresses on the impact resistance, the X-ray diffraction patterns, and the microstructure of 25S aluminum alloy. Many of the specimens were taken from actual propeller blades and others were cut from  $^{13}_{16}$ -inch rod furnished by the Aluminum Company of America.

The average impact resistances were found to be unchanged even after 288,000 cycles in a 0- to 33,400-pound-per-square-inch range that exceeded the fatigue limit and the original proof stress of the material. The X-ray diffraction patterns were unchanged as regards any indication of structural change resulting from the fatigue stressing of the alloy. Two structural conditions known as slip-plane precipitation and veining were observed. The service stresses were not responsible for the slip-plane precipitation and the endurance limit was not reduced by it. Veining could be made to disappear and reappear by alternate solution heat treatment and age hardening.

#### INTRODUCTION

During the course of its useful service life, any assembled structure, such as an airplane, is subjected to stresses varying greatly in their magnitude and nature. Although no single type of stress or simple combination of stresses can be considered most important in determining the duration of the service life, fatigue overshadows any other single factor under ordinary service conditions.

Recently Ravilly (reference 1) and Cazaud and Persoz (reference 2) have reviewed the various theories concerning the mechanism of the fatigue process. Study of the possible detrimental effect on the properties of a metal of continued fatigue stressing short of failure has, however, received scant attention. Attention has rather been directed to the determination of the number of applications of stress of known magnitude required to produce failure and the conditions that favor local overstressing, and therefore premature failure.

The possibility of detrimental changes occurring in the structure of the aluminum alloys used for aircraft propellers was made the subject of a group of tests conducted at the National Bureau of Standards laboratories in 1938. If damage did result, some short-time test to detect it was sought. In order to decide on a suitable testing procedure, two general assumptions regarding the nature of fatigue damage were made:

(a) The important changes occur throughout a large portion of the stressed body and are essentially dependent on the stress history of the body as a whole.

(b) The regions of extreme damage by repeated stresses are highly localized; hence, unless the effective cross section is reduced by the presence of cracks already formed, the use of any physical test whose results represent an averaging process throughout a large column of metal must necessarily show little if any correlation with fatigue damage.

Inasmuch as the main body of the work was devoted to (a), notch and corrosion effects were excluded as far as possible.

The subject was considered in three phases, each constituting a part of the paper. Part I, by J. A. Kies, deals with a possible lowering of the impact resistance of the metal after prolonged fatigue testing. Part II, by the same author, considered the use of X-ray diffraction patterns as a method of detecting fatigue damage prior to the actual failure of the member. Part III, by Kies and G. W. Quick, reports on slip-plane precipitation and veining.

Although no positive evidence of detrimental changes was obtained, the results are presented for the light they throw on the problem and the use of short-time methods to detect damage done to a metal by fatigue.

# I. IMPACT RESISTANCE OF 25S ALUMINUM ALLOY BEFORE AND AFTER FATIGUE STRESS-ING

Apparently little has been done on the subject of the possible lowering of impact resistance of aluminum alloys. K. Honda (reference 3) used the percentage decrease in impact resistance of plain carbon steels under repeated impact as an indication of the degree of fatigue damage. F. Oshiba (reference 4) showed a close correlation between the growth of fatigue cracks and the degree of fatigue in the case of plain carbon steels sub-

jected to repeated impacts and later (reference 5) reported a similar relation in the case of annealed plain carbon steel specimens of the rotating-beam type. In all cases, a small decrease in impact resistance was believed to precede the beginning of visible fatigue

In 1933 Moore and Wishart (reference 6) reported that lowering of the tensile strength resulted from stresses in excess of the fatigue limit provided that the number of cycles was sufficiently great. A value of 1,400,000 cycles was sufficient to establish fatigue limits in this manner for five steels, one brass, and monel metal. Although the number of reversals necessary for duralumin was not determined, it was thought to exceed the foregoing value. Earlier work by Davidenkow and Schewandin (reference 7) on annealed 0.15 to 0.20 percent carbon steel showed immediate lowering of the static breaking strength of notched specimens stressed above the fatigue limit in a rotating-beam machine and then broken by bending while immersed in liquid air. Brittle fractures permitted the course of the fatigue crack to be followed. It was found that, in specimens normally expected to fail after 700,000 cycles of the stress employed, the first visible evidence of damage appeared after 300,000 cycles. This result was considered to indicate that a general weakening in the material preceded the beginning of the crack. Neither the method of Davidenkow and Schewandin nor that of Moore and Wishart has led to any important application. It would appear, however, that sufficient evidence of important general changes exists to justify serious consideration of a program of impact testing of fatigued material. Presumably lowered impact re-

sistance in such material might be due to one of two causes, the presence of incipient cracks or to a cold-worked condition of the metal.

One of the most highly stressed members of any air-craft structure is the propeller. Despite the fact that no evidence of impending failure may be observed in the frequent inspections made of the structure, it is customary to place an arbitrary limit on the number

of hours of service permitted for any propeller. Because of such important considerations, an aluminum alloy in wide use for propeller blades was selected as the first material for study. Many of the specimens were taken from propeller blades in the possession of the National Bureau of Standards. Others were cut from 13/16-inch rod of 25S alloy furnished especially for the purpose by the Aluminum Company of America. The composition of the rod according to information furnished by the manufacturer, was:

Copper	4. 4 percent.
Manganese	0.76 percent.
Silicon	0. 82 percent.
Iron	0. 43 percent.
Aluminum	Remainder.

The heat treatment of aluminum alloy 25S recommended by the Aluminum Company of America (reference 8) consists in three steps:

- (a) Soaking at 960° to 980° F. The time required for this treatment depends on the load and on the nature of the heating bath.
- (b) Quenching in cold water, the temper designation then being 25SW.
- (c) Aging for 12 hours at 285° to 295° F., the temper designation being 25ST.

The tensile properties of the alloy 25ST rod as determined at the National Bureau of Standards were as follows:

TABLE I.—TENSILE PROPERTIES OF HEAT-TREATED ALUMINUM ALLOY, 25ST

Ultimate tensile strength	60, 600
True ultimate stress 1	
Nominal stress at fracture 2	52, 100
True stress at fracture 3	89, 300
Proportional limit 4	22, 300
Proof stress at 0.2 percent set	32, 300

- Maximum load divided by the actual cross-sectional area.
  Load at fracture divided by the initial cross-sectional area.
  Load at fracture divided by the actual cross-sectional area.
  Estimated stress when stress-strain curves departed from the slope of the modulus

The fatigue limit of 25ST as determined on the rotating-beam machine (R. R. Moore type) and based on 500,000,000 cycles of stress without failure, is given as  $\pm 15,000$  pounds per square inch (reference 8).

#### 1. TEST SPECIMENS AND MACHINES

The Haigh axial-loading fatigue-testing machine was used to stress in fatigue the material to be tested later for impact resistance. Figure 1 shows the fatigue



FIGURE 1.—Dimensions of the Haigh fatigue specimens used for the 13,000 pounds per-square-inch stress amplitude.

specimen used for the 13,000 pounds per square inch range, and figure 2 shows the specimens for the ranges higher than 13,000 pounds per square inch. Specimens machined from the fatigue bars were tested for impact resistance by the Charpy method and by the Luerssen-Greene torsion-impact method (reference 9).

Figure 3 (a, b, c,) gives the dimensions and form of the Charpy specimens, two different notch depths being used (0.005 and 0.039 inch). Figure 4 shows the torsion-impact specimens. The reproducibility of results obtained with the Charpy machine exceeded that obtained with the torsion-impact machine for the purpose of this work.

If W is the average impact energy indicated and  $\Delta W$  is the smallest difference of energy that can be read directly or estimated from the indicating mechanism, then

 $\frac{\Delta W}{W} = \frac{0.007 \text{ for } 0.005\text{-inch-notch depth}}{\text{(Charpy test)}}$ 

 $\frac{\Delta W}{W} = \frac{0.017 \text{ for } 0.039 \text{-inch-notch depth}}{\text{(Charpy test)}},$ 

and

 $\frac{\Delta W}{W} = 0.04$  (torsion impact).

These results are without regard to natural errors caused by faulty manipulation or to energy losses in the machines. In the case of the torsion tests, the impact value was meaningless when premature striking of the horns on the specimen holder occurred. In obvious cases of this kind, the test results were discarded.

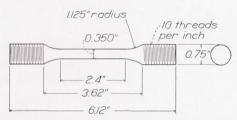
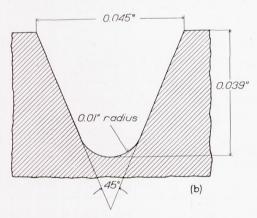


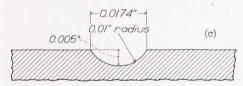
FIGURE 2.—Dimensions of the Haigh fatigue specimens used for stress amplitudes greater than 13,000 pounds per square inch.



(a) Machined from the Haigh fatigue specimens.



(b) Contour of the Charpy notch 0.039 inch deep.



(c) Contour of the special Charpy notch.

FIGURE .—Dimensions and form of Charpy type impact specimens.

The preliminary results showed that the percentage deviation from the mean for torsion impact tests was less than that for Charpy specimens with the 0.039-inch

notch depth but more than for Charpy specimens having notches 0.005 inch deep.

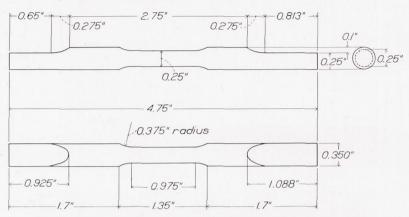


FIGURE 4.—Torsion impact test specimen machined from Haigh fatigue specimens.

#### TEMPERATURES FOR IMPACT TESTS

In the attempt to find test conditions that would yield the least ambiguous results, Charpy specimens comparable with those to be cut from Haigh fatigue specimens, were broken at various temperatures ranging from  $-190^{\circ}$  C. to  $99^{\circ}$  C. As is shown in figure 5, the smallest average percentage deviation from the mean was obtained at room temperature. Ample time for carefully centering the specimens may be the reason for the better uniformity at this temperature. The best obtainable agreement among check specimens was thought to be least obscuring to a possible change in impact resistance either with regard to scatter or to average value. On the other hand, low temperatures representing extremes encountered

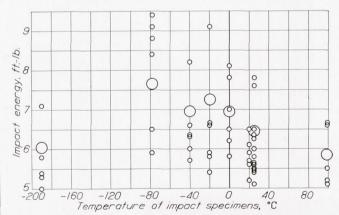


FIGURE 5.—Results of impact tests on Charpy type specimens of 25ST bar stock at various temperatures. The specimens are shown in figures 3(a) and 3(b). Each large circle indicates the midpoint of the scatter for one temperature.

in service could not be neglected and, accordingly, it was decided to make impact tests on fatigued material both at room temperature and at  $-78^{\circ}$  C.

### SIGNIFICANCE OF THE FIBERED APPEARANCE OF LONGITUDINAL SECTIONS OF 25ST ROD

A photograph (fig. 6) shows grains elongated in a direction parallel with the rod axis, but cross sections (fig. 7) show no preferential grain alinement. The

elongated grain structure is also quite prominent in 25ST rolled plate, and a definite difference in impact resistance is known to exist for specimens cut in the parallel and the transverse directions.

Preferred crystal orientation in varying degrees generally accompanies fibering. X-ray diffraction patterns can sometimes show the presence of preferred orientation. A diffraction pattern that showed definite preferred crystal orientation might therefore indicate indirectly the presence of a possible directional weakness in impact resistance.

X-ray diffraction patterns of the reflection type were obtained with a view toward devising a nondestructive

impact specimens cut longitudinally from the center of the 25ST rod need not have their notches oriented uniformly with respect to any certain rod diameter.

#### THE EFFECT OF SMALL VARIATIONS IN NOTCH DEPTH

Before the specimens were tested, the notch depths of the impact specimens were measured by a Hilger measuring microscope. Measurements were made at both ends of each notch. The average values plotted against the impact resistance showed no consistent relation between impact values and notch depth for the variations measured, that is, from -0.0015 to 0.0015 inch deviation from the nominal value. Larger

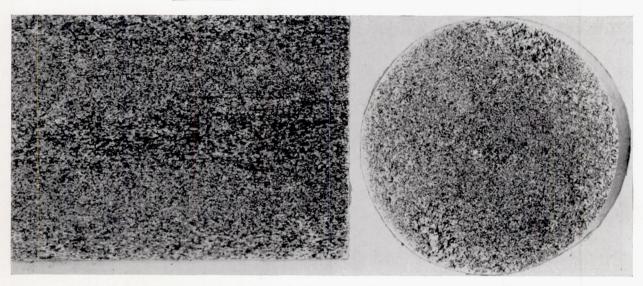


Figure 6.—Longitudinal section through 25ST rod showing elongation of the grains in direction of the rod axis,  $\times$  3.

FIGURE 7.—Transverse section through 25ST rod showing no preferential elongation of the grains in any direction, such as shown infigure  $6, \times 3$ .

method of inspection for propeller blades. The X-ray beam was allowed to fall on the etched specimen at grazing incidence and at right angles to the fibered direction. The specimen was moved in such a manner that the orientation of the fibered axis was constant with respect to the beam and to the film. In this way a large number of crystals was made to contribute to the pattern. No evidence of preferred orientation was found in the 25ST rod, in rolled 25ST plate known to have directional variation in impact resistance, or in a commercial duralumin rod, all of which showed a fibered appearance when polished and etched.

X-ray transmission patterns were obtained using thin etched slices of the 25ST rod. Transverse and longitudinal sections were studied, both stationary and moving. The section surface was kept in the same plane and no rotation about the direction of the beam was permitted. No preferred orientation was found.

It was concluded that X-rays are not suitable for the detection of directional variations in impact resistance in thick structural members such as aluminum-alloy propeller blades. Figure 7 shows no fibering transverse to the rod axis; consequently, it was decided that

variations, such as from 0.005 inch to 0.039 inch, do show marked effect on impact resistance.

#### TEST PROCEDURE AND RESULTS

Table II lists the various repeated stress treatments given the Haigh specimens.

TABLE II.—FATIGUE STRESSING OF SPECIMENS IN HAIGH AXIAL LOADING MACHINE AT ROOM TEMPERATURE PRIOR TO IMPACT

[The frequency was 2,400 cycles per minute in all cases]

		Stress	limits	Stress		
Mode of repeated stressing	Mean stress (lb./sq. in.)	Maxi- mum (lb./sq. in.)	Mini- mum (lb./sq. in.)	range (lb./sq. in.)	Cycles of stress	Remarks
a	3, 659 20, 900	10, 150 37, 600	-2,850 4,200	13, 000 33, 400	25, 000, 000 288, 000	Unbroken. Do.
C	20, 900	37, 600	4, 200	33, 400	153,000 to 480,000	Broken.
d	20, 900	38, 200	3,600	34, 600	144, 000	Unbroken.
e	20, 900	38, 200	3, 600	34,600	288, 000	Do.
f	20, 900	38, 200	3, 600	34, 600	192,000 to 276,000	Broken.
a. r	0	0	0	0	0	As received.

Specimens, machined from the reduced portions of the Haigh bars after stressing, were tested in impact, with the results shown in table III. It is evident that no significant embrittlement was induced by any of the repeated stress treatment, however damaging it may have been insofar as fatigue was concerned.

TABLE III.—IMPACT RESISTANCE OF 25ST ALUMINUM ALLOY PREVIOUSLY SUBJECTED TO FATIGUE STRESS

Impact test method	Num- ber of impact speci- mens broken	Temperature of test (° C.)	Mode of pre- vious fatigue stress 1	Notch depth (in.)	Average impact energy (ftlb.)	Average deviation from mean (percent)	Total width of scatter (ftlb.)
	( 18	25	a.r.	0. 039	6.0	8. 3	2. 6
	20	25	(a)	. 039	5. 7	9. 3	2.0
Charpy	{ 20	25	a.r.	. 005	14.6	5. 1	3. 4
	14 5	25	(b)	. 005	14.7	7.5	4.0
	5	25	(c)	. 005	15. 1	7.3	2.7
Torsion	ſ 18	25	a.r.	None	54. 0	7. 2	28.0
1 018(011	2	25	(a)	None	57.0	7.0	8.0
	[ 15	-78	a.r.	0.005	16.7	6.0	5. 2
	8	-78	(d)	. 005	16. 9	10.0	5. 4
Charpy	4	-78	(e)	. 005	15.7	7.6	2. 5
Charpy	3	-78	(f)	. 005	18.8	3. 2	1.8
	6	-78	a.r.	. 039	8.0	15.0	3. 5
	( 6	-78	(d)	. 039	6.5	21.5	4. 3

<sup>1</sup> See table II for mode of previous fatigue stress.

The distribution in impact energy of the individual specimens previously subjected to repeated stress was closely similar to that for the specimens machined from

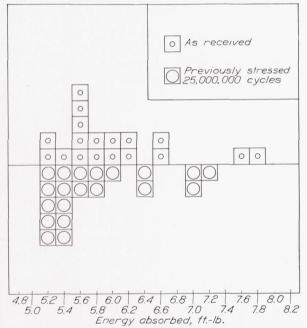


FIGURE 8.—Distribution of Charpy specimens according to energy absorbed in impact fracture. No significant change is found to result from prestressing 25,000,000 cycles in the range from 2,850 pounds per square inch compression to 10,150 pounds per square inch tension. Cyclic rate, 2,400 per minute; Charpy specimens were as shown in figures 3(a) and 3(b); each square represents one impact test at room temperature.

material in the "as received" condition. Figures 8, 9, 10, and 11 compare the distributions with the corresponding ones for unstressed material.

#### DISCUSSION OF RESULTS

In mode a (table II) of repeated stress treatment, it is reasonably certain that the specimens were stressed in a safe range. Assuming a sine wave for the stress-time relation, one can show that the original proof stress (at 0.2 percent set) was exceeded during approximately 24 percent of the cycle in modes b and c and during approximately 27 percent of the cycle in modes d, e, and f.

The decrease in specimen diameter during the repeated stress (modes b, c, d, e, and f) averaged  $0.43 \pm$ 

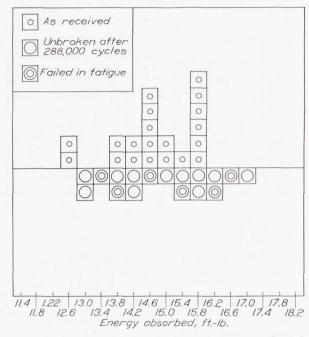


FIGURE 9.—Distribution of Charpy specimens according to energy absorbed. No significant change in impact resistance is found to result from repeated stresses in the unsafe range. The maximum tension exceeded the original proof stress of the alloy (25ST). Stress range, 4,200 to 37,600 pounds per square inch in tension; cyclic rate, 2,400 per minute; Charpy specimens were as shown in figures 3(a) and 3(c); impact testing temperature, 25° C.; each square represents one impact test.

0.03 percent with a corresponding increase in stress of  $0.86\pm0.06$  percent. This change is smaller than the uncertainty in adjusting the load in the Haigh machine. It also illustrates the fact that the amount of *general* cold work was very slight.

#### CONCLUSIONS

The average impact resistances, as measured by Charpy tests at 25° C. and at —78° C. and by torsion-impact tests at 25° C., were unchanged by any of the following conditions of prior fatigue stressing at 25° C.:

25,000,000 cycles of safe stress, the total range being 13,000 pounds per square inch, part of which was in compression. Torsion and Charpy impact tests were made at room temperatures.

288,000 cycles in a 33,400 pounds per square inch range that exceeded the fatigue limit and the original proof stress of the material. All the stress was in tension, and the number of cycles exceeded half the greatest number required to cause failure in five similar specimens. Charpy impact tests were made at room temperature.

Various numbers of cycles (153,000 to 480,000) causing fatigue failure in a stress range of 33,400 pounds

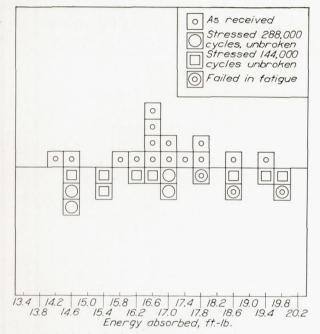


FIGURE 10.—Distribution of Charpy specimens according to energy absorbed at  $-78^{\circ}$  C. Special notch; no embrittlement was found to occur as a result of stressing alloy 25ST for various numbers of cycles in the stress range 3,600 to 38,200 pounds per square inch in tension; impact specimens of the type shown in figures 3(a) and 3(c) were used; each square represents one test.

per square inch, all of which was in tension. Charpy impact tests were made at room temperature.

288,000 cycles in a 34,600 pounds per square inchrange. The Charpy tests were made at  $-78^{\circ}$  C.

Various numbers of cycles (192,000 to 276,000) causing fatigue failure in a stress range of 34,600 pounds per square inch. Charpy impact tests were made at  $-78^{\circ}$  C.

#### II. X-RAY DIFFRACTION PATTERNS OF 25S ALU-MINUM ALLOY BEFORE AND AFTER FATIGUE STRESSING

The great advantage of a nondestructive test for determining the extent to which metals may have been damaged by fatigue scarcely needs explanation. In addition to possessing the advantage of not injuring the metal tested, the X-ray method can be applied to small areas on the surface of fatigue specimens or structures under investigation. It is well known that severe fatigue damage is usually highly localized and that the geometric forms of the specimens and the method of stress application frequently induce the beginning of

the fatigue cracks in or near the surface at locations of relatively high stress concentration.

The successful application of X-ray diffraction to the study of metals that may have been damaged by fatigue stressing depends on whether or not the changes detectable by X-ray diffraction are the significant ones in fatigue. The question of whether or not the X-ray diffraction method may be of use for the purpose is still controversial, and the correct testing procedure is by no means standardized. The aim in such studies, of course, has not been to detect actual cracks, for which other methods of inspection are in common use, but to tell whether or not the formation of a crack is imminent.

The essential details of what happens in the early stages of a fatigue failure, before a crack has been started, are obscure; hence, X-ray diffraction studies are as yet empirical as far as correlation with fatigue is concerned. It is important to interpret correctly the

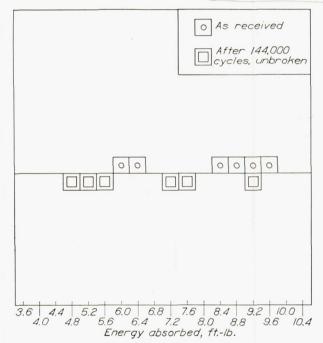


FIGURE 11.—Distribution of Charpy type specimens according to energy absorbed at -78° C. Standard notch; no embrittlement was found to occur as a result of stressing alloy 25ST for 144,000 cycles of stress in the range 3,600 to 38,200 pounds per square inch in tension.

changes in diffraction pattern in terms of crystal structure in order to learn the limitations and the possibilities of the method.

According to published results, changes in an X-ray diffraction pattern occurring as a result of fatigue stressing may all be attributed to a condition of cold working in the metal. The various effects of cold working on metallic crystals have been classified (references 10 and 11), together with the changes in X-ray pattern associated with them, as follows: Block dislocation with negligible lattice distortion, extended block dislocation, lattice distortion, and preferred orientation. It should be understood that all these effects accompany the break-down of metallic crystals by cold work and

are expected to occur simultaneously to varying degrees.

Block dislocation with negligible lattice distortion connotes the fragmentation of crystals into smaller units or blocks that tilt slightly with respect to the orientation of the parent grain. The diffraction pattern obtained with monochromatic radiation shows the diffraction spots of the original grains smeared along the circumference of the diffraction (Debye) rings to which they belong. No radial widening of the rings occurs. Continuous radiation gives a diffraction pattern characterized by radial asterism that varies with the X-ray method used.

Extreme amounts of cold work may result in continued fragmentation and extended dislocation of the crystal. The diffraction spots of the parent structure are then indistinguishable. With monochromatic radiation a continuous diffraction (Debye) ring is formed and the smaller the crystallites are, the wider and more diffuse is the diffraction ring up to a certain limit. Continuous radiation produces a general fogging of the diffraction pattern.

Elastic deformation produces changes in the lattice constants. The changes vary from grain to grain in a polycrystalline material, and even in a single crystal a given lattice spacing may vary from point to point. Elastic bending cannot occur without producing some lattice distortion (reference 11); the resulting curved reflecting crystal planes yield qualitatively the same diffraction effects as the small dislocations described.

Preferred orientation results from a relatively high degree of cold working and occurs by readjustment of crystalline fragments in certain preferred directions that depend on the crystalline habit of the parent crystal and the mode of stress application causing fragmentation. The preferred orientation encountered in cold-rolled sheet ordinarily requires a fairly high degree of general plastic deformation of the metal before its presence can be shown by X-ray diffraction patterns.

Some of the recent studies by X-ray methods (references 12, 13, and 14) of fatigue damage have been accomplished by locking the test specimen in the same position for each exposure to X-rays and observing the changes in individual diffraction spots from time to time after increasing numbers of stress cycles. Very considerable difficulty of reproducing exactly the same system of spots is encountered.

A number of the earlier attempts at correlating changes in X-ray diffraction patterns with the progress of fatigue have been reviewed and summarized by Barrett (reference 15). Recently Gough and Wood (reference 16) associated changes in X-ray patterns with (a) dislocation of the grain into components that vary in tilt up to 2° from the orientation of the parent grain; (b) the formation of crystallites or fragments approximately  $10^{-4}$  to  $10^{-5}$  centimeters in diameter; and (c) the presence of lattice distortion.

Gough and Wood found progressive changes in the diffraction patterns obtained from specimens of normalized 0.12 percent carbon steel subjected to repeated stresses in the unsafe range. Some of the specimens were subjected to alternating torsion and others to alternating axial tensile stress.

Erich Martin (reference 17) found that changes in X-ray pattern during repeated stress depend on the previous condition of the metal. For instance, a coldworked steel containing 0.17 percent C, 0.71 percent Si, and 0.90 percent Mn gave a back reflection Debye diffraction ring in which the Ka<sub>1</sub> and Ka<sub>2</sub> doublet was not resolved. After 1,000 repeated impacts in an unsafe range, the specimen showed the doublet nicely resolved, whereas the initially blurred doublet from a heat-treated rimming steel remained blurred after similar repeated impact. Tests by Gough (reference 18), however, indicated that resolution of the Ka doublet does not always occur as a result of fatigue stressing cold-worked steel. For instance, in the case of a cold-rolled lowcarbon steel, the diffraction patterns showed a uniform diffuse diffraction ring before and after repeated stressing.

Wever and Möller (reference 13) decided that there is too much variation from place to place on a fatigue specimen to make X-ray patterns taken at random positions dependable criterions of damage. Accordingly, they followed the practice of etching a small spot to localize the fatigue failure. In addition, the specimen was always locked in the same position with respect to the beam and the film so that the etched spot was always irradiated. The same diffraction spots could then be studied on each of a series of films. Wever and Möller believed that safe and unsafe stress could be differentiated in this way, although the question of whether the dividing line was indicated with high accuracy was still unsettled. They emphasized again the need for a progressive study of each specimen. Barrett (reference 19) disagreed, however, with Wever and Möller on their ability to foretell fatigue failure from X-ray data.

Later Wever, Hempel, and Möller (reference 14) confirmed the previous work of Wever and Möller on annealed steel and concluded that fatigue damage could also be detected in the case of cold-worked steel. They depended on the progressive study of diffraction spots for stresses above the fatigue limit. Möller and Hempel (reference 12) published a detailed account of their examination of annealed low-carbon steel fatigue specimens and, although the test seems to indicate that plastic deformation is the condition to which X-ray patterns are sensitive, Möller and Hempel believed that the observed changes could differentiate between safe and unsafe stresses.

All the published work to date indicates that no single X-ray examination is sufficient to show whether or not a specimen or a service member has been dam-

aged by prior fatigue stressing. In fact, it would appear that more than two such examinations are necessary. A comparison must be made from time to time with previous photograms; even then the possibility of showing fatigue damage has not been demonstrated except for a very few metals and no demonstration in the case of a service member was found in the literature.

#### MATERIALS, TESTS, AND RESULTS

In view of the incompleteness of Barrett's work on the aluminum alloy 25ST, attempts were made to find evidence of damage to this material after it had been subjected to fatigue stress for various periods. The first specimens were cut from an airplane propeller of 25ST alloy, which was re-heat-treated to the 25SW condition. A composition typical of this alloy has been given in Part I of this report. Material in the W, or quenched, condition was first selected because of the lower stress necessary to produce plastic deformation in such material. The surface of the reduced section of a Krouse fatigue specimen, which is of the cantilever type, 2 inches long and 0.25 inch in diameter with a central section reduced to 0.185 inch in diameter, was etched to a depth of about 0.002 inch. The specimen was keved in such a position that the same spot could be used for each successive photogram. An X-ray beam, iron Ka radiation, was directed at the specimen at grazing incidence and the diffraction patterns shown in figure 12 (a) to 12 (e) were obtained. The diffraction patterns obtained are therefore representative of the same specimen in the course of a series of fatigue-stress applications. Figure 12 shows the photograms after each treatment as follows:

- (a) Before being fatigue-stressed.
- (b)  $\pm 12,000$  lb./sq. in. for  $11.9 \times 10^6$  cycles.
- (c)  $\pm 14,500$  lb./sq. in. for  $1.8 \times 10^6$  cycles.
- (d)  $\pm 16,000$  lb./sq. in. for  $1.5 \times 10^6$  cycles.
- (e)  $\pm 17,000$  lb./sq. in. for  $1.4 \times 10^6$  cycles, plus  $8.7 \times 10^6$  cycles at  $\pm 18,000$  lb./sq. in.

No changes of known significance in the diffraction pattern after stressing at successively increasing amplitudes are apparent. After additional stressing at  $\pm 18,000$  pounds per square inch nominal stress amplitude, the specimen was broken.

Further studies were made with the Haigh axial-loading fatigue machine on specimens cut from <sup>13</sup>/<sub>16</sub>-inch rod of the fully heat-treated aluminum alloy 25 ST, but no progressive changes in the X-ray diffraction patterns that can be considered as significant for foretelling failure in a fatigued member have been found. The grain size of the material was small enough to produce a large number of spots on the diffraction patterns so that one difficulty encountered by Barrett (reference 16)

while studying this alloy, namely, that of too few spots, was not a consideration.

The specimens were carefully machined with a narrow-nosed sharp tool, finishing with four or five cuts 0.0025 inch deep, followed by two cuts 0.001 inch deep. The reduced sections and fillets were then polished longitudinally with 1G emery paper and aloxite. Four trial specimens were etched to various depths and examined to determine if the X-ray patterns were influenced by the polishing process. Molybdenum radiation and a collimating system consisting of two holes in lead, one millimeter in diameter and 5.8 centimeters apart, were used throughout. The group of diffraction patterns in figure 13 (a) to 13 (d) are in order of the depth to which the surfaces of four specimens were etched: 0, 0.0005, 0.0025, and 0.0075 inch, respectively. It will be noted that the spots are equally sharp in all cases. It was concluded that etching was unnecessary.

The next step was a progressive study of a Haigh bar fatigue-stressed in a tensile range sufficient to cause failure, as follows:

Speci- men 4 <sub>H</sub> 4 <sub>A</sub>	Stress range 4,200 to 37,600 lb./sq. in	. tension	
Date stressed	Cycles	Date of X-ray	Figure
2/23/38 4/29/38 5/26/38	288,000	4/27/38 4/28/38 4/29/38 6/2/38	14 (a) 14 (b) 14 (c) 14 (d)

Figure 14 (b), (c), and (d) were all obtained on the specimen after locking it in a position so that approximately the same area was repeatedly exposed to the beam, although the settings were not accurate enough to give exactly the same diffraction spot pattern each time. It was concluded, however, that the general nature of the diffraction pattern remained unchanged during the period in which the specimen was subjected to fatigue stressing. The fact that crystal fragmentation or dislocation occurred close to the crack may be seen by the drop in intensity of the higher-order reflections in figure 14 (e). A slight diminution in relative intensity of the outer rings with respect to the inner ones seems to have occurred in figure 14 (d). Figure 14 (e) shows a somewhat greater loss in intensity in the outer rings. The significance of this result is somewhat doubtful in the absence of confirmatory evidence.

A similar result was obtained on Haigh specimen (3H5B) stressed in tension from 0 to 33,400 lb./sq. in. The record follows:

Date stressed	Cycles	Date of X-ray	Remarks
Not stressed	36,000	5/2/38 5/6/38 6/9/38	No change. Do.



(a) Before being fatigue-stressed, etched to 0.002 in



(b) After  $11.9 \times 10^6$  cycles at  $\pm 12,000$  lb./sq. in.



(c) After 1.8×106 additional cycles at ±14.500 lb./sq. in.

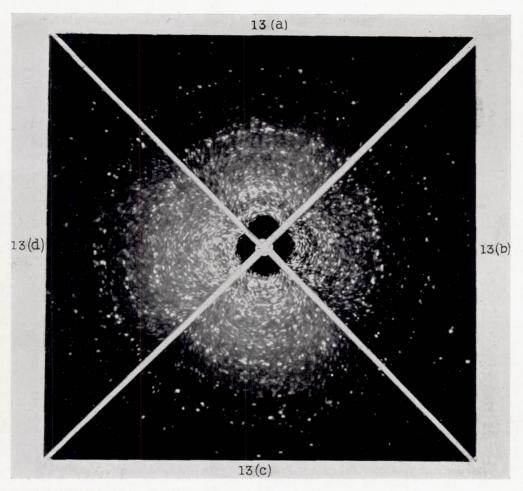


(d) After  $1.5\times10^6$  additional cycles at  $\pm16,000$  lb./sq. in.



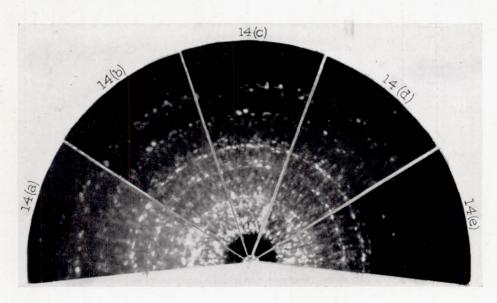
(e) After 1.4×10 $^{\rm d}$  additional cycles at  $\pm 17,000$  lb./sq. in. plus  $8.7\times 10^{\rm d}$  cycles at  $\pm 18,000$  lb./sq. in.

Figure 12.—Grazing incidence patterns made by reflections of iron radiation from the same spot on an etched fatigue specimen of 258W after successive treatments. No extended crystal fragmentation is indicated although the continuous portion of some of the rings in (e) show a suggestion of a slight increase in intensity.



(a) 0 inch. (b) 0.005 inch. (e) 0.0025 inch. (d) 0.0075 inch.

FIGURE 13.—Diffraction patterns showing equal sharpness of spots after polishing on aloxite paper and etching to various depths.



- (a) and (b) Taken at the end of 288,000 cycles, different areas being exposed to the X-ray beam.
  (c) After 36,000 additional cycles on the same spot as (b).
  (d) The same spot after 79,200 additional cycles, resulting in fatigue failure. The crack formed several centimeters from the X-rayed spot.
  (e) Diffraction pattern at the fatigue crack after failure. The spot exposed to the X-ray beam was near the origin of the crack.

Figure 14.—Progressive X-ray study during the last stages of fatigue of alloy 25ST stressed in the range 4,200 to 37,600 pounds per square inch in tension by the Haigh method.

Likewise, another Haigh specimen (3H2A) stressed repeatedly in tension from 3,600 to 38,200 pounds per square inch showed no general changes to have occurred between the 144,000 cycle and the 180,000 cycle. All three of these specimens were stressed in the unsafe range.

Since considerable time had elapsed in some cases between the end of the period of stressing and the exposure to the X-rays, a check was made on the possibility of an obscuring effect by recovery at room temperature. No effect suggestive of a recovery in the previous tests was found, however, in material stressed 1 cycle between 0 and 30,000 pounds per square inch or between 0 and 26,000 pounds per square inch in tension.

These conditions were first noticed during the examination of an airplane propeller blade made of the aluminum alloy 25ST that had failed in service. The same structural features were subsequently detected in other propeller blades. Although there was no a priori reason for regarding the material with suspicion because of the presence of these features, their discovery in a structural member that had failed in service prompted a study of their origin and probable significance.

#### SLIP-PLANE PRECIPITATION

The usual appearance of the condition of slip-plane precipitation, as found in specimens cut from a propeller blade of the 25ST aluminum alloy, is shown by figures

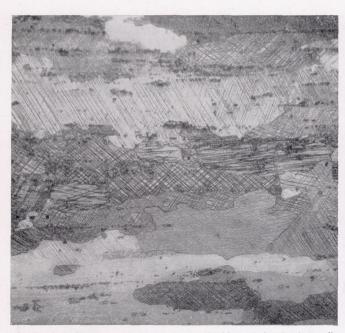


Figure 15.—Slip-plane precipitation in a propeller blade of the aluminum alloy 25ST. The fact that the criss-cross lines change direction at grain boundaries proves that the markings are not scratches.  $\times$  100.

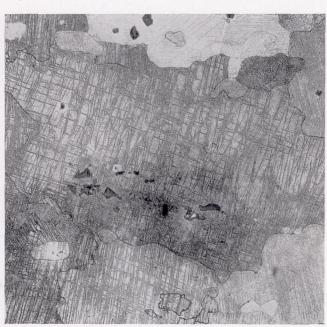


FIGURE 16.—Slip-plane precipitation in a propeller blade of the aluminum alloy 25ST. The offsetting of the lines in some localities suggests shearing in the direction of lines crossing the offset ones.

Exposures of 9 hours were used and about 1 hour elapsed between stressing and the beginning of the first exposure.

#### CONCLUSIONS

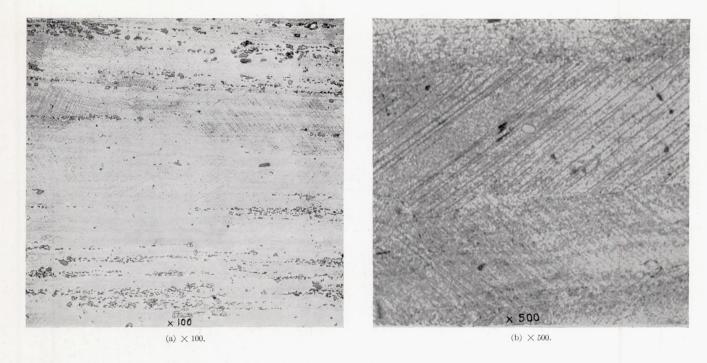
The results of the foregoing studies on the aluminum alloy 25S failed to show any difference in the X-ray diffraction patterns that could be considered significant of structural changes resulting from prior fatigue stressing of the alloy.

#### III. SLIP-PLANE PRECIPITATION AND VEINING IN 25S ALUMINUM ALLOY BEFORE AND AFTER FATIGUE STRESSING

The terms "slip-plane precipitation" and "veining" apply to quite unrelated structural conditions. The appropriateness of these designations, especially the first one, will appear as the results of the work are described.

15 and 16. Many of the individual grains are criss-crossed by intersecting groups or families of parallel lines. Any doubt as to the confusion of these lines with scratches produced in polishing the metallographic specimen is removed by noting the abrupt change in direction of the lines from one grain to another. The fact that the individual lines are not continuous but that each one consists of a series of discrete particles is revealed by the series of micrographs at successively increasing magnification shown in figure 17 (a) to 17 (d).

The hardening, by heat treatment, of the aluminum copper alloys, of which the 25S alloy is representative, consists of three steps. The first is a solution treatment during which the alloy is heated to a temperature sufficiently high to permit the diffusion into solid solution of the various structural constituents of the alloy; this treatment is immediately followed by quenching in water to retain this structural condition at ordi-



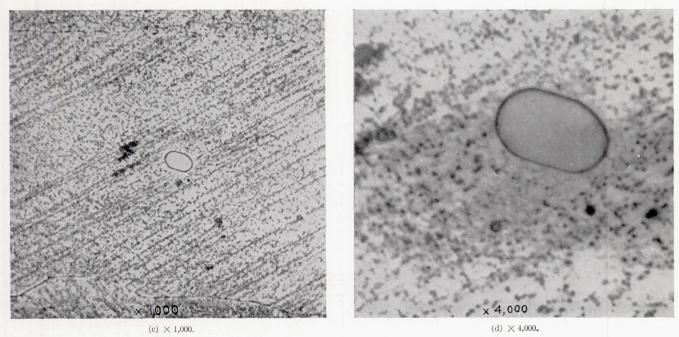


FIGURE 17.—Slip-plane precipitation at the same location on a polished and etched surface of aluminum alloy 25ST at different magnifications. The lines of (a) are resolved at higher magnifications into rows of particles. Etchant, 1/2 percent HF.

13

nary temperature. The third step in treatment, aging or reheating at a slightly elevated temperature or simple aging at room temperature in case of a few alloys, permits the precipitation of the excess of the constituents held in solid solution in the matrix of the unstable or quenched alloy in the form of tiny particles dispersed throughout the matrix of the alloy (reference 20). (Details of the heat treatment of the 25S alloy are given in Part I of this report.) Fink and Smith (reference 21) have shown that indications of the precipitation in the aluminum-copper alloys may be detected after as little as one-half hour of aging time following the solution treatment and that the particle size increases as the time of aging is extended. The 25S alloy is essentially an aluminum-copper alloy, as is shown by the table on page 2.

The crystal lattice of the 25S alloy behaves during plastic deformation in the same manner as any other face-centered cubic lattice, that is, deformation occurs by slippage on octahedral planes. The micrographs given in figure 17 indicate that the term "slip-plane precipitation" correctly describes the phenomenon. Observations by Wassermann (reference 22) on a similar condition in the structure of an aluminum-copper alloy are in confirmation of those reported here.

A further step in the study of slip-plane precipitation consisted in the identification of the crystal planes on which the precipitated phase was localized.<sup>1</sup> identification was done on some of the large crystals found in a section of a propeller blade. Figure 18, a micrograph of one of these crystals, shows the directions of the slip-plane precipitation. The directions were measured with respect to reference scratches drawn on the polished surface. The crystal orientation was determined in the usual way from the X-ray diffraction pattern obtained by the back-reflection method. Stereographic projections (reference 23) of the octahedral planes are represented by small circles called poles in figure 19. The loci of normals to the rows of precipitated particles of figure 18 are projected as straight lines in figure 19. These lines were prepared in their proper relation to the reference scratches and the polished plane of the specimen. In all the cases examined the octahedral poles fell on the normals to the slip lines. This result means that the octahedral planes were in such a position that their traces in the plane of polish coincided with all the directions taken by the rows of precipitated particles. No other single family of poles of simple indices would account for all the slip-line directions. Confirmatory evidence that only the

<sup>&</sup>lt;sup>1</sup> The valuable assistance of H. C. Vacher, Metallurgist, National Bureau of Standards, is acknowledged in this phase of the work.



of this crystal is shown in figure 19. The 100. alloy 25ST, X grain of the heat-treated aluminum in a large FIGURE 18.—Slip-plane precipitation

octahedral planes of the crystals are concerned in this structural change is obtained by counting the number of directions of the observed slip lines. If these lines

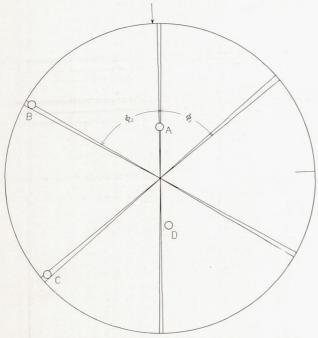


FIGURE 19.—A stereographic projection of octahedral poles, A, B, C, and D, as determined from a back-reflection diffraction pattern of the 25ST crystal shown in figure 18. The straight lines are projections of the loci of normals to the observed slip lines. All the observed slip lines can thus be accounted for by glide on three of the four octahedral planes,  $\Phi_1$ =48° to  $51\frac{1}{2}$ °;  $\Phi_2$ =58 $\frac{1}{2}$ ° to 62°. The angles between the normals to the various octahedral planes and the normal to the surface of polish are:  $\Theta_A$ =37°;  $\Theta_B$ =87 $\frac{1}{2}$ °;  $\Theta_C$ =87 $\frac{1}{2}$ °;  $\Theta_D$ =34°.

are the traces of only octahedral planes, then four is the maximum possible number of directions that may be marked out by the precipitated particles in any plane

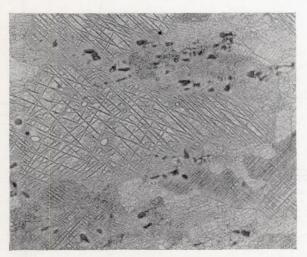


FIGURE 20.—Slip-plane precipitation in the heat-treated aluminum alloy 25ST. × 200. Note four families of lines in one grain; the maximum possible number if slip-plane precipitation is confined to octahedral planes.

of polish. The observations are in agreement with this rule. Figure 20 depicts a grain showing four families of slip planes and figure 21 is its stereographic repre-

sentation. Fewer directions than four are usually observed but no more than four were found in this investigation.

## PRODUCTION AND ELIMINATION OF SLIP-PLANE PRECIPITATION

A question that immediately arises concerning the structural condition is whether this change from the normal structure occurred during service, presumably as a result of service stresses, and whether the condition is detrimental to the material, particularly its resistance to fatigue stresses. In a study of the fatigue characteristics of telephone-cable sheath, Townsend (reference 24) concluded that structural changes occurred in antimony-lead alloy as a result of the service conditions, primarily of the stresses and that these changes had an important bearing on the subsequent behavior of the material as a whole. In that case, the change, which consisted essentially of the precipitation from solid solution of the antimony, was confined principally

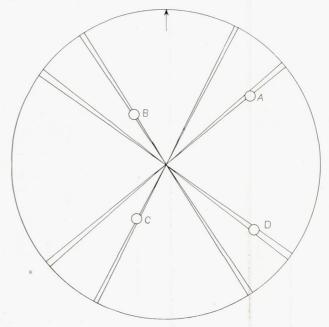


FIGURE 21.—A stereographic projection of octahedral poles, A, B, C, and D, of the aluminum lattice in a possible position such that each slip line of figure 20 is accounted for by glide on octahedral planes. The straight lines are projections of the loci of normals to the observed traces of slip planes and, as in figure 19, the maximum variation in direction among members of each set of slip lines is indicated on the drawing. The angles between the normals to the various octahedral planes and the normal to the surface of polish are:  $\Theta_{\Lambda} = 70^{\circ}$ ;  $\Theta_{D} = 42^{\circ}$ ;  $\Theta_{C} = 43^{\circ}$ ;  $\Theta_{D} = 70^{\circ}$ .

to the regions of the grain boundaries. The possibility of a similar occurrence in other cases, such as the present one, with the precipitated phase located on certain favored crystallographic planes, should not be overlooked.

In addition to an estimation of the probability of a damaging condition arising from the presence of slipplane precipitation in the 25ST alloy, it is also desirable to know the circumstances under which this condition may and may not be produced.

In the numerous attempts made, it was found impossible to produce slip-plane precipitation in either 25SW or 25ST alloy in quantities visible in the microscope, by merely deforming the material, either elastically or plastically. Fatigue stresses of various kinds have also been tried and negative results obtained. Also all attempts to produce slip-plane precipitation in fully

and then aging it at elevated temperatures, some of the modes of deformation being:

- (a) Compressive deformation under a Brinell ball, under a forging press, and under a forging hammer.
- (b) Slow tensile deformation. Figure 23 shows a photograph of slip-plane precipitation on a longitudinal section developed by stretching a tensile specimen of

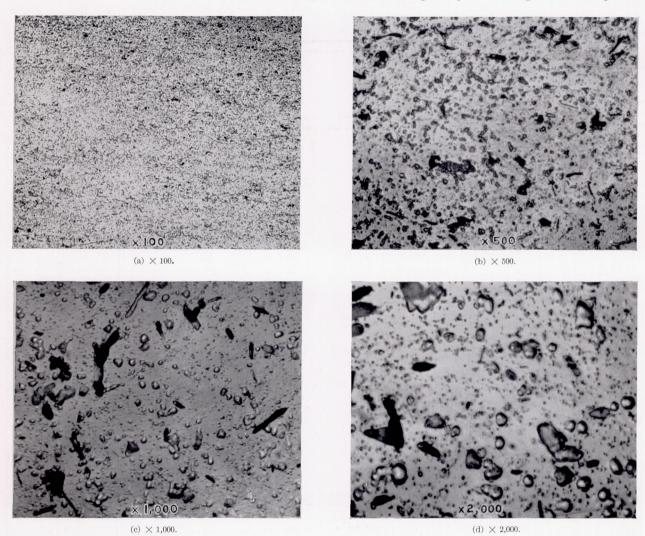


FIGURE 22.—The same location of a polished and etched surface of aluminum alloy 25S at different magnifications. All the micrographs were taken at the end of a sequence of treatments consisting in a slow cooling from 525° C. to room temperature, plastic bending, and heating for 18 hours at 143° C. in air. Note that the precipitated particles are not arranged in rows as in figure 17. Presumably the failure to produce the familiar slip-plane precipitation is explained by the fact that insufficient quantities of material remained in solid solution after the annealing process to permit further precipitation on slip planes.

annealed 25S alloy by deforming it in various ways and by subsequent aging for various periods at 143° C. failed to develop this structural condition. Figure 22 shows the appearance of 25S alloy annealed, bent, and aged 18 hours at 143° C. It is evident that, although precipitation from the matrix of microstructural particles occurred presumably during the annealing process, no tendency toward localized precipitation on slip planes of the crystals was observed.

No slip-plane precipitation was found in any specimen of 25S alloy in the quenched condition. This structural feature was produced in the alloy, however, by deforming the quenched material in various ways quenched 25S alloy then aging at 143° C. for 17 hours. The slip lines are fairly straight.

- (c) Impact by the Charpy and the tensile impact methods. Figure 24 is a photograph of a longitudinal section of a broken tensile impact specimen subsequently aged at 143° C. for 17 hours. The slip lines are crooked in this case.
- (d) Plastic bending, one side of the specimen being in compression and the opposite in tension. A curious property of this type of deformation is the great difference of abundance of slip-plane precipitation found in various parts of specimens. Figure 25 shows a macrograph of two bars of 25S alloy, both of which were bent

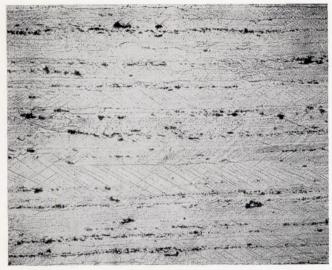


FIGURE 23.—Slip-plane precipitation developed by slow tensile deformation of a specimen of the quenched aluminum alloy 25SW followed by aging at 143° C. for 17 hours. × 100.



FIGURE 24.—Slip-plane precipitation resulting from deformation of a specimen of the quenched aluminum alloy 23SW in tensile impact followed by aging at 143° C. for 17 hours. × 100.

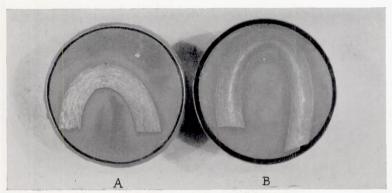


FIGURE 25.—Macrograph of sections of specimens of the aluminum alloy 25S bent while in the quenched condition. A, as quenched; B, aged after quenching. × 1. Note the curved light streak in the center of B marking the region of greatest density of slip-plane precipitation. A showed no slip-plane precipitation and etched more uniformly.

cold while in the quenched condition. Bar B was subsequently age-hardened whereas A was not. The white streak along the neutral axis of B marks the region of greatest abundance of slip-plane precipitation and the zone widens in the portions of the specimen least deformed. No slip-plane precipitation was found in A. The uniform etching of the specimen in this case is noteworthy.

(e) Fatigue failure. One end of a failed fatigue specimen, after fracture, was split longitudinally perpendicular to the fatigue fracture. Figures 26 and 27 show the pair of "imaged" surfaces thus obtained. The portion aged at elevated temperature, after being fractured by fatigue stressing, shows slip-plane precipitation whereas the imaged portion was practically free from it.

It is not always necessary to deform plastically the quenched 25S alloy by mechanical means in order to produce slip-plane precipitation. Quenching and agehardening, in themselves, are sometimes sufficient, although the amount of slip-plane precipitation obtained is ordinarily much less than if the quenched alloy has

been mechanically deformed. Figure 20 illustrates the condition found near one corner of a rectangular block quenched and age-hardened without any mechanical working. It is not to be concluded, however, that absolutely no plastic deformation occurred because it has been shown by Kempf, Hopkins, and Ivanso (reference 25) that the quenching process leaves residual stresses. These stresses may approach the yield point of the material for extreme quenching conditions.

Perhaps the most important of all the observations on the production of slip-plane precipitation are those made on fatigue specimens of the quenched alloy 25SW and of

the quenched-and-aged alloy 25ST. The results of the observations, stated briefly, are:

No slip-plane precipitation was found in any fatigued specimen of quenched 25S alloy, stressed or not stressed, until after an aging treatment at elevated temperatures.

Fatigue stressing followed by aging, however, did not invariably result in slip-plane precipitation in the quenched alloy 25SW. Rotating-beam specimens of quenched 25S alloy stressed in fatigue, but not to complete failure, at various amplitudes from ±14,000 to ±20,000 pounds per square inch showed no slip-plane precipitation even after aging at 143° C. to 146° C. for 18 hours. Aging, at 165° C. to 171° C. for 16 hours, of specimens stressed above and below the fatigue limit, did produce some slip-plane precipitation, but no special importance could be attached to this result since unstressed control specimens cut from the material adjacent to the fatigue specimens also developed slip-plane precipitation when aged under similar conditions.

Specimens of quenched-and-aged alloy 25ST, stressed in the Haigh axial-loading fatigue machine for 25,000,-

000 cycles in the range from 2,850 pounds per square inch in compression to 10,150 pounds per square inch in tension were not visibly different in structure from companion, unstressed material in respect to the amounts of slip-plane precipitation developed by equal aging treatments. Figure 28 (a) to 28 (f) show the close parallelism between the unstressed and stressed material as aging proceeds.

FIGURE 26.—Slip-plane precipitation in the plastically deformed portion of a Moore rotating-beam fatigue specimen. The 258 specimen was broken by repeated stress while in the quenched condition. The portion adjacent to the fracture was split longitudinally; one portion was aged 6 hours at 145° C. and used for this figure; the other portion was used for figure 27.  $\times$  100.

### PRACTICAL SIGNIFICANCE OF SLIP-PLANE PRECIPITATION

Of most importance in connection with determining the practical significance of slip-plane precipitation is its relation to fatigue resistance, especially as to whether it is an indication of weakness in fatigue.

It was found that 25ST propeller-blade material could be requenched and aged according to standard commercial practice with only a slight decrease in Vickers hardness number. After this reheat treatment, the material contained only comparatively small amounts of slip-plane precipitation, as is shown by figures 29 and 30. The results of determinations of the fatigue strength of the alloy as received and of the requenched-and-aged material revealed no marked difference between the two but rather a slight superiority, if anything, in favor of the material containing much slip-plane precipitation. Figure 31 shows the results obtained for these two materials, all the fatigue

tests having been made on the same Krouse cantilever rotating-beam machine. The fatigue limits based on  $10^8$  cycles of stress without failure, were  $\pm 18,500$  and  $\pm 19,000$  pounds per square inch for the requenched-and-aged alloy and for the as-received material, respectively. It is therefore evident that, although slipplane precipitation occurs in the quenched alloy 25SW, only as a result of deformation followed by aging, the



FIGURE 27.—This surface has been separated from the one shown in figure 26 by a longitudinal cut. The portion shown here was not aged at an elevated temperature. No slip-plane precipitation is in evidence although plastic deformation was present to about the same degree as in the portion shown in figure 26.  $\times$  100.

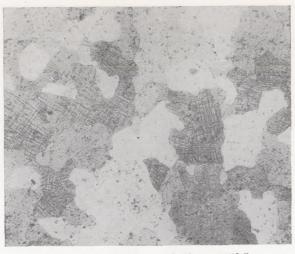
existence of this structural condition in the material does not constitute a dependable criterion of prior fatigue stressing. Material that had been intentionally subjected to fatigue stressing often did not exhibit this structural condition on being aged after being stressed.

#### VEINING

A network structure termed "veining" that occurs within crystals is prevalent in service structures of the quenched-and-aged alloy 25ST and, although there is no a priori reason to be suspicious of it, a study was made to show whether or not this phenomenon can be used as an indication of damage by fatigue stressing. Experiments showed that quenching and aging are the important factors necessary in causing its extinction and recurrence, respectively. The following characteristics of veining were those to which attention was directed in making the observations and experiments.



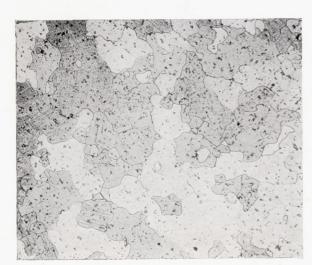
(a) Not stressed in fatigue, aged 10 hours at 159° C.



(c) Not stressed in fatigue, aged 18 hours at 159° C.



(e) Not stressed in fatigue, aged 4 hours at 145° C. and 30½ hours at 159° C.



(b) Fatigued, 25,000,000 stress cycles, and aged as in (a).



(d) Fatigued, 25,000,000 stress cycles and aged as in (c).



(f) Fatigued, 25,000,000 stress cycles and aged as in (e).

FIGURE 28.—The close parallelism in the development of slip-plane precipitation in specimens of the fully heat-treated aluminum alloy 25ST as received from the maker and after fatigue stressing, 25,000,000 cycles of axial loading in the range from -2,850 to 10,150 pounds per square inch. × 100.

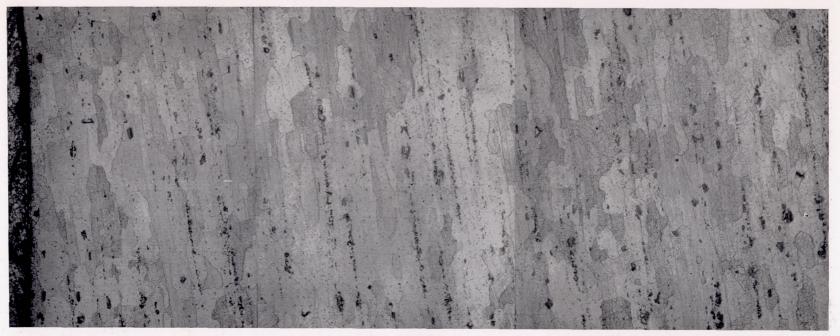


FIGURE 29.—A section through a 25ST aluminum alloy propeller blade after quenching and aging to produce nominal full hardness. Note the absence of slip-plane precipitation. × 100.



FIGURE 30.—A section through the same propeller blade shown in figure 29, as received. Much more slip-plane precipitation is in evidence than in the quenched and age-hardened sample shown in figure 29.  $\times$  100.

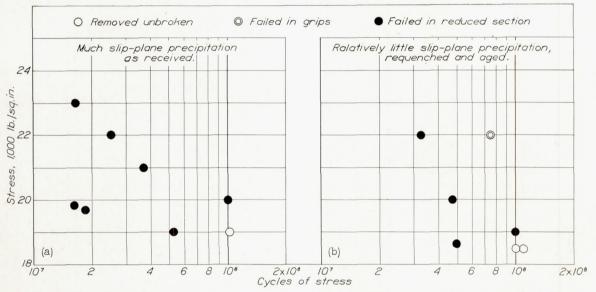
The veins are continuous and give the impression of being cell walls. Figure 32 (a) and (d) show typical examples.

If the veining is an indication of subdivision into sub grains, it is evident that the orientation is essentially the same throughout the limits of the mother grain. Figure 32 (c) shows slip planes crossing the veins and apparently not deviated by them.

Veining is eliminated, insofar as its visibility is concerned, by the solution heat treatment. Figure 32 (b) shows approximately the same region as figure 32 (a) after heating at 521° C. and quenching in cold water.

Veining can be suppressed and restored repeatedly by alternately quenching and aging, which is in general Cold plastic deformation of 25S aluminum alloy in the quenched condition in appropriate amounts followed by aging at the temperature usually employed, namely, about 143° C. for 15 to 18 hours, produces copious amounts of slip-plane precipitation, but if the material is initially in the age-hardened condition, very little slip-plane precipitation is obtained under like aging conditions. Deformation without subsequent aging at elevated temperatures produces no slip-plane precipitation

Small amounts of slip-plane precipitation normally result from the sequence of solution treatment and age-hardening. Regions near the periphery, and especially the corners of angular specimens where quench-



(a) Much slip-plane precipitation, as received.

(b) Relatively little slip-plane precipitation, requenched and aged.

Figure 31.—Results of Krouse rotating-beam cantilever fatigue tests on two samples of aluminum alloy 25ST, one containing much more slip-plane precipitation than the other, as shown in figures 29 and 30.

agreement with the findings of Northcott (reference 26) in the case of other metals. A specimen cut from an airplane propeller blade was alternately quenched and aged. The treatments and results of the subsequent metallographic examinations are tabulated in table IV, the observations being carried out in each case on the same grain.

TABLE IV.—EFFECT OF HEAT TREATMENT ON VEINING IN THE ALUMINUM ALLOY, 25S

Treatment	Structure
Solution heat treatment, quenched in water from 521° C. (970° F.).	No veining.
Aged 6 hours at 145° C	Veining present, figure 32 (d). No veining, figure 32 (e).
Aged 6 hours at 145° C	Veining present, figure 32 (f).

#### CONCLUSIONS

The general conclusions of this study of two structural conditions observed in the 25S aluminum alloy are summarized as follows:

The correctness of the term "slip-plane precipitation" has been established.

Service stresses are not responsible for slip-plane precipitation in 25ST propeller blades.

ing strains are supposedly greatest, are the regions of greatest density of slip-plane precipitation.

The use of slip-plane precipitation as a reliable indication of damage by fatigue stress is not promising because of the frequent presence of this structural feature in 25ST stock whose fatigue resistance is known to be unimpaired.

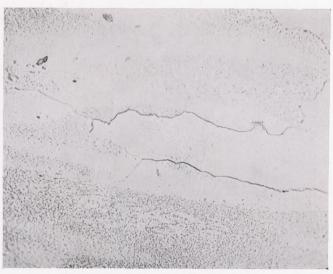
The endurance limit of the quenched-and-aged aluminum alloy 25ST, is not reduced by the presence of slip-plane precipitation in the amounts ordinarily found in the material.

The structural condition termed "veining" can be made to disappear and reappear repeatedly by alternately subjecting the material to the solution heat treatment and to age-hardening, without any further stress treatment; hence, it is improbable that veining can be used to betray damage in this alloy by fatigue stresses.

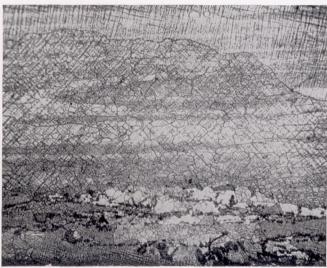
NATIONAL BUREAU OF STANDARDS WASHINGTON, D. C., October 1938.



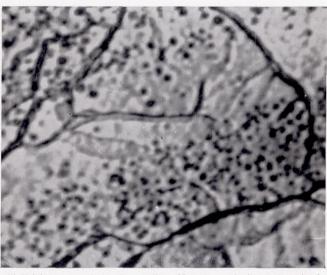
(a) Veining in a specimen, as received. X 500.



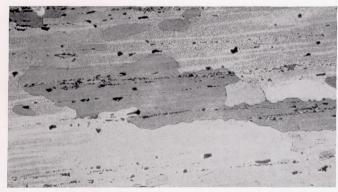
(b) Approximately the same region as in (a), after solution heat treatment.  $\times$  500.



(c) Showing both slip-plane precipitation and veining. No apparent deviation in the direction of the slip lines occurs on crossing the veins and indicates that the various subdivisions are in approximately the same orientation throughout. × 200.



(d) Veining in quenched and aged alloy. Note the continuity of the veins in contrast to the particle nature of slip-plane precipitation. No veining was found in this specimen in the previous, or quenched, condition. Aging treatment, 6 hours at 145° C.  $\times$  3,600.



(e) Large dark grain free from veining. Same grain shown in (d) after a quench from 525° C, followed by a light polish and etch.  $\times 100$ .



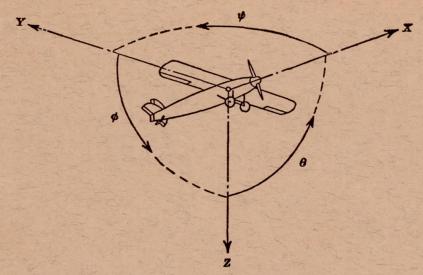
(f) Same grain as in (d) and (e) after having been aged at 146° C, for 6 hours, repolished, and etched. A very thin layer was removed by polish so as to make identification of the grain certain. The veining is restored although it is not so distinct as before. Even if the same system of veins had been restored, the fact could not have been ascertained owing to the removal of the surface layers in polishing. × 100.

FIGURE 32.—Veining and slip-plane precipitation in the heat-treated aluminum alloy 25ST as received and after different treatments.

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Positive directions of axes and angles (forces and moments) are shown by arrows

1000	Axis			Moment about axis			Angle	9	Velocities	
+ + -	Designation	Sym- bol	Force (parallel to axis) symbol	Designation	Sym- bol	Positive direction	Designa- tion	Sym- bol	Linear (compo- nent along axis)	Angular
The state of the s	Longitudinal Lateral Normal	X Y Z	X Y Z	Rolling Pitching Yawing	L M N	$ \begin{array}{c} Y \longrightarrow Z \\ Z \longrightarrow X \\ X \longrightarrow Y \end{array} $	Roll Pitch Yaw	φ θ ψ	u v w	p q r

Absolute coefficients of moment

$$C_i = \frac{L}{qbS}$$
 (rolling)

$$C_m = \frac{M}{qcS}$$
 (pitching)

$$C_n = \frac{N}{qbS}$$
 (yawing)

Angle of set of control surface (relative to neutral position), δ. (Indicate surface by proper subscript.)

#### 4. PROPELLER SYMBOLS

- D, Diameter
- Geometric pitch p,
- p/D, V',  $V_s$ , Pitch ratio
- Inflow velocity
- Slipstream velocity
- Thrust, absolute coefficient  $C_T = \frac{T}{\rho n^2 D^4}$ T,
- Torque, absolute coefficient  $C_Q = \frac{Q}{\rho n^2 D^5}$ Q,
- Power, absolute coefficient  $C_P = \frac{P}{\rho n^3 D^5}$
- Speed-power coefficient =  $\sqrt[5]{\frac{\overline{\rho V^5}}{Pn^2}}$  $C_s$
- Efficiency η,
- Revolutions per second, r.p.s. n,
- Effective helix angle= $\tan^{-1}\left(\frac{V}{2\pi rn}\right)$

#### 5. NUMERICAL RELATIONS

- 1 hp.=76.04 kg-m/s=550 ft-lb./sec.
- 1 metric horsepower=1.0132 hp.
- 1 m.p.h.=0.4470 m.p.s.
- 1 m.p.s.=2.2369 m.p.h.

- 1 lb.=0.4536 kg.
- 1 kg=2.2046 lb.
- 1 mi.=1,609.35 m=5,280 ft.
- 1 m=3.2808 ft.

